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E/RAE
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REPORT No: MET.7

E/RAE-MET 7

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ROYAL AIRCRAFT ESTABLISHMENT

Farnborough, Hants.

ATI No. 10137

THE RELATION BETWEEN THE TENSILE PROPERTIES AND THE CONSTITUTION OF ALUMINIUM-RICH ALUMINIUM-MAGNESIUM-MANGANESE-ZINC ALLOYS

by

A. H. ROSS

and

B. W. MOTT, M.A.

Air Documents Division, T-2
AMC, Wright Field
Microfilm No.

RC-347F 10137

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U.D.C. No. 669.715.721 : 539.42.015

R.A.E. Report No. Met. 7

August, 1946

ROYAL AIRCRAFT ESTABLISHMENT, FARNBOROUGH

The relation between the tensile properties and
constitution of aluminium-rich
aluminium-magnesium-manganese-zinc alloys

by

A.H. Ross
and
B.W. Mott, M.A.

R.A.E. Ref: Mat.M7790/BWM/171
M. o S. Ref: Res. Mat. 785/R.D. Mat. /M

SUMMARY

Introduction

In view of the high tensile properties obtained^{1,2} on aluminium alloys of nominal composition 3% of magnesium, 1% of manganese and 7% of zinc in the wrought and heat treated condition, it was considered desirable to ascertain the effects of wide variations in the magnesium and zinc contents on the tensile properties and microstructure.

Range of Investigation

Thirtyone alloys were prepared with magnesium contents in the range 0.86 to 4.55%, manganese contents from 0.96 to 1.12% and zinc contents from 2.88 to 12.7%. Cast billets were forged to $\frac{1}{8}$ inch square bar and tensile tests were made in the fully heat treated condition. The materials were microscopically examined in the cast, as forged and forged and fully heat treated conditions.

Conclusions

For a given percentage of magnesium, the values of 0.1% proof stress and ultimate tensile strength increased steadily to a maximum with increasing zinc content while the elongation value decreased. Similarly for a given percentage of zinc, increases in proof and ultimate stresses were obtained for an increase in the magnesium content. The relationship between the diamond pyramid hardness and tensile strength of the alloys was linear. A possible theoretical relationship between tensile strength and composition is discussed.

Values for the 0.1% proof stress, ultimate stress and elongation for the alloy containing 2% of magnesium and 11% of zinc in the fully heat treated condition were 37.0 tons/in.², 40.1 tons/in.² and 6% respectively and for the alloy with 4% of magnesium and 11% of zinc, 40.4 tons/in.², 44.5 tons/in.² and 2% respectively. Microscopical

examination of the alloys in various conditions showed that the constitution of the alloys was very similar to that of high purity materials.

Further Developments

Further work is in hand to include a full investigation of the tensile, fatigue and stress-corrosion properties of forged, extruded and rolled material in an alloy containing 2 to 3% of magnesium, 1% of manganese and 10 to 11% of zinc. The effects of additions of chromium up to 0.5% on the tensile, stress-corrosion and working properties of a selected series of wrought alloys of high tensile strength are being investigated.

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U.D.C. No. 669.715.721 : 539.42.015

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The relation between the tensile properties and
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M.o S. Ref: Res.Mat.785/R.D.Mat.M

R. Introduction

It is well known that aluminium-rich alloys containing magnesium and zinc possess high tensile properties in the wrought and heat treated condition. The strength properties obtained at R.A.E. on alloys containing about 3% of magnesium, 1% of manganese and 7% of zinc and on an alloy with 3% of magnesium, 1% of manganese, 7% of zinc and 0.5% of chromium, have been reported.^{1,2,3,4} The constitution of the Al-Mg-Mn-Zn system has been investigated by Dr. Hume-Rothery and his collaborators^{5,6,7,8} and the results of microscopical examination of alloys of this type have been reported by R.A.E.^{2,9,10,11}

In view of the very high tensile properties which have been obtained on the 3:1:7 alloy both with and without additions of chromium, it was desirable to ascertain the effects of variations in the magnesium and zinc contents on the tensile properties and microstructure. The present report gives the results of tensile tests of material in the forged and heat treated condition of a series of thirty one alloys in the range 0 to 5% magnesium, 0 to 12% zinc, and containing 1% of manganese together with small quantities of iron and silicon as impurities. The microstructure of the alloys was examined in the cast, forged, and forged and heat treated conditions.

Since the initiation of this work, Specification D.T.D. 363A was issued in March, 1944 to cover bars for machining and extruded sections in material which conformed to the following range of composition.

copper,	%	up to 3.0
magnesium,	%	up to 4.0
manganese,	%	up to 1.0
iron,	%	up to 0.6
silicon,	%	up to 0.6
nickel,		nil
zinc,	%	4.0 to 8.0
titanium,	%	up to 0.3
chromium,	%	up to 1.0
aluminium,		the remainder.

The minimum tensile properties required by the Specification are a 0.1% proof stress of 33.0 tons/in.², a tensile strength of 38.0 tons/in.² and an elongation value of 5%, for sections of thickness between $\frac{3}{8}$ inch and 6 inches.

Experimental Procedure

A series of alloys was prepared with compositions falling within the ranges

magnesium, %	1.0 to 5.0
zinc, %	3.0 to 12.0
manganese, %	1.0

iron and silicon being present as impurities only. The basic materials used in preparing these alloys were all of good commercial quality. The chemical analyses are given in Table I and a list of the nominal compositions of the alloys investigated is given in Table II.

Chilled billets of about 4 pounds weight were prepared, melting being carried out in a covered Salamander crucible in an electric resistance furnace. No difficulty was experienced in preparing the alloys provided that the necessary precautions were taken to avoid undue loss of magnesium and zinc. Full details of the preparation, melting, casting and subsequent treatments of the alloys are given in Appendix I. The forgeability of the materials throughout the whole range of the alloys investigated was satisfactory and remarkably consistent.

Chemical composition

The compositions of the alloys investigated are given in Tables IIIa, IIIb, IIIc, and IIId. The magnesium content of the alloys is within the range 0.86 to 4.55%, the manganese content within the range 0.96 to 1.12%, and the zinc content within the range 2.88 to 12.7%. An addition of from 0.04 to 0.06% of titanium was made to all alloys for grain refinement and improved working properties. Iron and silicon were present as impurities only in all the alloys, the average contents being about 0.33% and 0.14% respectively.

Heat Treatment Experiments

Previous work^{2,4} having shown that soaking for one hour at 495°C. was satisfactory for alloys of this type, initial solution heat treatment experiments were made at this temperature on alloys reference nos. 3:1:5, 3:1:6, 3:1:8, 3:1:12, 4:1:6 and 5:1:12, where the numbers refer to the nominal percentage of magnesium, manganese and zinc respectively.

Although bars 3:1:5, 3:1:6, 3:1:8 and 4:1:6 appeared to be satisfactory, superficial examination of bars 3:1:12 and 5:1:12 after this treatment showed the presence of blisters and other signs of overheating. Microscopical examination of specimens of the 3:1:12 and 5:1:12 alloys solution treated at 475°C. showed that formation of films and pools of the ternary eutectic, ($\alpha + \text{Al}_2\text{Mg}_3\text{Zn}_3$), had occurred which indicated that this temperature was above the solidus temperature for each alloy. Solution treatment at 450°C. however resulted in partial solution of the ternary phase and a globularisation of the MgZn_2 and any undissolved ternary, without the formation of further separations of ternary constituent which would cause embrittlement of the material. Cooling curves were plotted for alloys 3:1:8 and 4:1:11. In these, a small arrest was noted in alloy 3:1:8 at 472°C. and a more marked arrest was evident in alloy 4:1:11 at 476°C.

Small pieces of bars 1:1:3, 1:1:4, 3:1:3 and 2:1:5 were solution heat treated at 520°C. for one hour, quenched in cold water, and artificially aged for 18 hours at 120°C. Microscopical examination revealed no sign of overheating in any of these samples and no indication was obtained from Brinell hardness tests that the tensile properties of the materials treated at 520°C. would be superior to those after treatment at 495°C. It was evident, however, that as the zinc content was increased it was necessary to lower the solution heat treatment temperature to avoid overheating, and it was decided that all alloys with a zinc content of over 9% or a total of magnesium plus zinc of over 12% should be solution treated at a lower temperature. A temperature of 470°C. proved to be satisfactory for bars 1:1:10, 1:1:11, 2:1:10, 2:1:11, 3:1:10, 3:1:11, 4:1:9, 4:1:10 and 4:1:11.

Low properties obtained on a 5:1:12 bar solution treated at 450°C., were due to the incomplete solution of the $Al_2Mg_3Zn_7$ phase at this temperature. All the remaining bars were solution treated at 495°C. for one hour, followed by quenching in cold water.

Tests made on alloys of this type to determine the effects of varying the conditions of ageing had shown that the maximum values of proof and ultimate tensile stresses were obtained by ageing at a temperature of from 120°C. to 140°C. for a period of about 18 hours.^{2,4} With increasing temperature of ageing above 140°C., the values of proof and maximum stresses decreased and the percentage elongation increased. The ageing treatment adopted for all bars of the series with a view to developing the optimum tensile properties in each alloy was 18 hours at 120°C.

Discussion of tensile properties

Owing to the small quantity of material available, in general only one tensile test was made on each alloy, but in a few cases, where there was sufficient material, duplicate tests were made. Also for bars 3:1:9 and 4:1:8 in which, in the initial test, fracture occurred in the end thread, a second test piece was taken from the remaining part of the original bar. Full details of the results of mechanical testing are given in Tables IIIa, IIIb, IIIc and IIId. For alloys containing nominally 1%, 2%, 3% and 4% of magnesium, the values of maximum stress were plotted against actual percentage of zinc in each case and are shown in Figures 1a, 1b, 1c, and 1d respectively. (The Tables show that the average variation in magnesium content from the nominal value is about 0.2% for each series.)

Examination of Figures 1a, 1b, 1c and 1d shows that, provided that the temperature is controlled to permit maximum solution of constituents without exceeding the solidus of the alloy, in any given series with constant magnesium content, the proof stress and ultimate tensile strength values increase steadily to a maximum as the zinc is increased while the elongation value decreases.

For given percentages of zinc, the values of proof stress and ultimate tensile strength show an increase of approximately 7% as the magnesium content is increased from 1 to 2%, of about 3% for an increase from 2 to 3% of magnesium and about 1% for an increase from 3 to 4% of magnesium.

Although the values of proof stress and ultimate tensile strength in the curves show a tendency to further slight increase for the 4:1:11 alloy, the percentage elongation has decreased to 2%. It was considered that alloys of such low ductility in the wrought and heat treated condition have no practical value for structural parts of aircraft and there would be no purpose in increasing the magnesium and zinc contents still further.

In order to correlate the values of ultimate tensile strength obtained for all alloys with the respective magnesium and zinc contents, the position of the alloys on the ternary diagram were plotted in terms of the atomic percentages of free magnesium and of zinc as shown in Fig.1e. The atomic percentage of free magnesium was calculated after allowing for the amount required to form Mg_2Si with all the silicon present in the alloy. The number adjacent to each point in Fig.1e indicates the value of maximum stress in $tons/in.^2$ obtained on forged material of the respective composition as given in Tables IIIa, IIIb, IIIc and IIId. The dotted lines are tentatively suggested as representing the shape of the curves of equal tensile strength.

Bungardt and Schaitberger have indicated¹² that, for alloys within a certain range of composition as given by the boundaries of the $(\alpha + T)$ field in the ternary Al-Mg-Zn equilibrium diagram at the temperature of ageing, the $AlMg_2Zn_3$ phase is responsible for age hardening, whereas for alloys in the $(\alpha + MgZn_2)$ field, $MgZn_2$ is the hardening phase. The tensile strength obtained after certain fixed conditions of solution treatment and ageing will obviously be dependent on the amount of precipitation which has occurred. It is suggested that the precipitation after a given time of ageing may be related to the form of the solid solubility boundary on the diagram, i.e. the line between the α and $(\alpha + T)$ regions and that between the α and $(\alpha + MgZn_2)$ regions which, according to Little, Raynor and Hume-Rothery, can be represented respectively by the equations

$$\frac{[Mg]}{[Zn]} = \text{constant}$$

$$\text{and} \quad \frac{[Mg]}{[Zn]^2} = \text{constant}$$

where $[Mg]$ and $[Zn]$ are the atomic percentages of free magnesium and of zinc respectively.

The tensile strength of alloys in the $(\alpha + T)$ and $(\alpha + MgZn_2)$ fields will then be functions of $[Mg]/[Zn]$ and $[Mg]/[Zn]^2$ respectively or alternatively alloys of equal tensile strength in a given phase field will give a constant value of the corresponding product. If a sufficiently large number of alloys were selected, this hypothesis could be checked by constructing the two sets of curves for which $[Mg]/[Zn]$ is constant and $[Mg]/[Zn]^2$ is constant, on the equilibrium diagram and seeing if they represent curves of equal tensile strength in the corresponding phase fields. In view of the small number of alloys investigated and the lack of accurate information on the equilibrium diagram at $120^\circ C.$, it was not possible to check these relationships.

Brinell hardness tests were made on the ends of the test pieces under conditions of loading $H_{10}/1000/15$ and the results are given in Tables IIIa, IIIb, IIIc, and IIId. Owing to the limited amount of material available, the values given are generally those obtained by single determinations only. Accordingly, small slices with two parallel faces were machined from the heads of the tensile pieces after testing and diamond pyramid hardness determinations were made under conditions of loading $HD/30$. The results given in the Tables are the average of ten determinations for each alloy.

In Fig.2, the average diamond pyramid hardness is plotted against the corresponding value of tensile strength. Within the limits of experimental error there is a linear relationship between the hardness and tensile strength over the range of alloys investigated.

Examination of Macrostructure

A section was machined from each of the chill cast billets for

examination of macrostructure. For the majority of the ingots the sections were taken halfway along the length but in some cases, the section was cut from a position just below the head of the billet. Some fine porosity, which varied with the alloy, was observed in the centre of the sections but in no case did the macrostructure suggest that the material would not forge satisfactorily.

The sections were etched in a solution containing 5% of hydrofluoric acid, 15% of hydrochloric acid and 20% of nitric acid. Examples of the structures obtained are given in Fig.3 which shows the sections from 3:1:10, 1:1:10, and 1:1:11 alloys. In most cast billets the grains were fine, uniform and equiaxed as in the 3:1:10 alloy. Slight columnar growth had occurred at the edges of some ingots, e.g. 1:1:10 alloy, but no excessive columnar growth was observed. In a few cases there was a ring of very fine grains on the edge of the ingot, then a band of coarser grains while the grain size in the centre was intermediate between that of the two outer regions (1:1:11 alloy).

Examination of microstructure

The microstructure of the cast alloys was examined in sections taken from the slices used for macro-examination and of the forged alloys in sections from the tensile test pieces. In the range of alloys investigated four constituents were identified in the microstructure. The manganese was present as Mn Al₆ and any iron occurring in the alloys as an impurity was considered as being in solution in the manganese aluminide. The magnesium and zinc formed the ternary constituent Al₂Mg₃Zn₃ in alloys of low alloy content but in those with higher percentages of magnesium and zinc, some MgZn₂ particles were observed. Small quantities of magnesium silicide were present but it was not certain that all the silicon contained in the alloys was in this form. The mode of occurrence and the etching characteristics of these phases were in general similar to those of the constituents in high purity alloys.^{5,6,7,8,11}

Cast Bars

No ternary phase was observed in the cast alloys containing only 1% of magnesium until the concentration of zinc was 10%. In alloys in the range 1:1:3 to 1:1:9, the microstructure contained separations of the iron-bearing manganese aluminide particles together with small quantities of magnesium silicide. The manganese constituent was present in a finely divided state and in some areas the particles were distributed so that the phase appeared to be present as a eutectic. A filigree form of (Fe-Mn-Al) was a common feature in alloys of low alloy content and an example of this is shown in Fig.4. Only small amounts of magnesium silicide were observed and none of the typical filigree form of this constituent, which is usually present in alloys containing magnesium and silicon, was found. It is possible that some of the silicon present in the alloys as an impurity was in solution in the manganese constituent. In the 1:1:10 and 1:1:11 alloys small quantities of the ternary constituent had precipitated in the grains and grain boundaries.

With a magnesium content of 2% and zinc contents up to 5%, the microstructures were similar to those of corresponding alloys with 1% of magnesium. In the 2:1:6 alloy and in alloys of higher zinc contents, small amounts of the ternary phase were visible. No appreciable separations of the eutectic (α +T) was observed in this range of alloys.

No ternary phase was present in the 3:1:3 alloy but it occurred as a fine precipitate in the 3:1:4 sample in an amount similar to that in which it was present in the 2:1:6 alloy. The precipitate was more profuse in the 3:1:5 alloy but no appreciable separations of resolvable eutectic

were observed until the zinc had been increased to 6% for a corresponding magnesium content of 3%. A typical field in the 3:1:6 alloy is illustrated in Fig.4 which shows the three constituents:- Mg_2Si (Fe-Mn-Al), and $Al_2Mg_3Zn_7$, after the section had been lightly etched in a 0.1% solution of sulphuric acid saturated with hydrogen sulphide. The fine precipitate of the ternary phase can be seen in Fig.4 both in the grain boundaries, (indicated by P_1) and in the grains, (indicated by P_2). A progressive increase in the amount of ternary phase was observed in the 3:1:8, 3:1:9, 3:1:10 and 3:1:11 alloys. The separations in the grain boundaries were generally larger and the eutectiferous form coarser with increase in zinc depending upon the particular cooling conditions for each ingot. In some alloys, the ternary constituent was present in a globularized form in the grains.

A more marked change of microstructure was observed in the 3:1:12 sample. The occurrence of the manganese aluminide and magnesium silicide particles was as in the alloys previously described, but a new magnesium-zinc phase, $MgZn_2$, was present. The $MgZn_2$ is similar in appearance to $Al_2Mg_3Zn_7$, but unlike the ternary phase, does not occur in a eutectiferous form. The $MgZn_2$ particles are often hexagonal in outline as shown in Figs.5 and 6, (5:1:12 alloy). The $Al_2Mg_3Zn_7$ phase had separated out in the grain boundaries as large pools of coarse ($\alpha+\beta$) and large quantities of the ternary had been precipitated in the grains as small needles as shown in Fig.6.

The microstructures of the 4:1:6, 4:1:8 and 4:1:9 alloys contained small quantities of $MgZn_2$ which appeared to be present only in areas in which marked segregation of constituents had occurred. The ternary eutectic was not so coarse as in the 3:1:12 alloy and more of this phase had separated out in the grains in the form of globularized pools, rather than the profuse fine precipitation as observed in the 3:1:12 alloy. In the 4:1:8 sample the magnesium silicide was present in the unusual form of prismatic particles which were often present in the larger pools of the ternary constituent, probably as the result of the thermal conditions during cooling. As would be expected the amounts of $MgZn_2$ were greater in the 4:1:10 and 4:1:11 alloys than in the others of this series but no appreciable coarsening of the structure was observed.

The structure of the 5:1:12 alloy is illustrated in Figs.5 and 6. The ($\alpha+\beta$) separations were coarse and the manganese aluminide particles were larger than in any of the other alloys. Appreciable quantities of the $MgZn_2$ phase were present and the typical prismatic shape of this constituent is clearly visible in Fig.6. The precipitated $Al_2Mg_3Zn_7$ was very acicular and profuse.

Treatment at 450°C., i.e. at the temperature used for soaking the billets prior to forging, resulted in the complete solution of the ternary phase in the cast alloys containing low percentages of magnesium and zinc. In some of the alloys in which the separations of $Al_2Mg_3Zn_7$ were somewhat coarse, only partial solution of this phase was brought about by soaking for one hour at this temperature. Neither the Mg_2Si nor the $MgZn_2$ phases were appreciably affected by treatment at 450°C.

Forged bars

A few of the alloys were microscopically examined after air cooling subsequent to the forging operation. All alloys had been broken down satisfactorily although some slight segregation of the manganese constituent was observed. Alloys with high percentages of magnesium and zinc contained quantities of both $MgZn_2$ and $Al_2Mg_3Zn_7$. Some of the ternary phase had separated out during air cooling after the forging operation and any large particles of $MgZn_2$ or of undissolved $Al_2Mg_3Zn_7$ were broken down by the forging treatment.

With the exception of those alloys of high zinc content that contained appreciable amounts of $MgZn_2$ in the cast condition, the microstructures of the forged bars heat treated at $495^\circ C$. were all similar to each other. Heat treatment at $495^\circ C$. for one hour resulted in the complete solution of the ternary $Al_2Mg_3Zn_7$ phase and subsequent ageing at $120^\circ C$. caused precipitation, the presence of which was revealed by etching the sections in various reagents. The etching reagents that gave the most satisfactory results were a 0.5% solution of hydrofluoric acid, a 25% solution of nitric acid at $70^\circ C$. and a 0.1% solution of sulphuric acid saturated with hydrogen sulphide. The amount and distribution of the precipitated particles varied with the alloy and was most profuse in those alloys containing higher percentages of magnesium and zinc. The size and distribution of the precipitated particles in the 4:1:6 alloy can be observed in Fig.7. The precipitation was not uniform but appeared to be more profuse in some grains than in others. Another feature observed was an apparent recrystallisation which had taken place as revealed in the large grain occupying the centre of the field of Fig.7. This was common to all the alloys but was more marked in those of higher alloy content. After etching the specimens in a solution containing hydrofluoric and nitric acids, examination showed an apparent duplex structure to a varying degree in most alloys. This feature has been observed previously in a 3:1:7 alloy.² Two distinct phases appeared to be present in the matrix which were distinguished by a difference in shade and also in the amount of precipitation which had occurred. The effect was most clearly observed after long periods of etching and is illustrated in Fig.8, (3:1:8 alloy).

No satisfactory explanation of this apparent duplex structure can be given but the following suggestions are made. The etching characteristics of a given grain may be dependent on its orientation so that variation in the depth of attack over the section would be expected. Alternatively, it is believed that microsegregation of constituents occurs not only from grain to grain but also within a single grain and this is borne out by the mode of separation of liquated constituents on overheating. Under such circumstances, it is possible that the composition of some areas will correspond to the $(\alpha + MgZn_2)$ phase region and others to the $(\alpha + T)$ phase region. Although the etching characteristics of the $Al_2Mg_3Zn_7$ and $MgZn_2$ phases are very similar, it is possible that those of regions in which $MgZn_2$ has precipitated would be different to those of areas containing precipitation of $Al_2Mg_3Zn_7$. Although it is doubtful if segregation occurs to such a marked extent as to provide a satisfactory explanation to the peculiar etching characteristics of the heat treated alloys, it would explain the observation that the duplex structure is more apparent in some alloys than in others.

Of the six alloys containing $MgZn_2$ in the cast state, those with composition 4:1:9, 4:1:10 and 4:1:11 were successfully solution treated at $470^\circ C$., and the 5:1:12 alloy at $450^\circ C$. Although no tensile test was made on material of composition 3:1:12 after solution treatment at $470^\circ C$., it is most probable that satisfactory tensile properties would be obtained by ageing after solution treatment at this temperature.

Sections from the 3:1:12, 4:1:8 and 5:1:12 alloys were examined after solution heat treatment at $495^\circ C$. The 3:1:12 and 5:1:12 samples showed marked overheating. Large quantities of $(\alpha + T)$ eutectic had separated out in the grain boundaries and small particles of $MgZn_2$ were observed in the grains. In some areas in the 5:1:12 alloy there was an almost continuous grain boundary network of ternary constituent as shown in Fig.9. (The section was deeply etched in a dilute solution of hydrofluoric acid.) Some fine intercrystalline cracking is indicated by the arrow in the photograph. The sample of 4:1:8 alloy was only slightly overheated at $495^\circ C$. The $MgZn_2$ particles were globularized and most of the $Al_2Mg_3Zn_7$ constituent was taken into solution. In some areas however in which micro-

segregation had occurred the ternary constituent had separated out as fine films in the grain boundaries and as small rings and pools in the grains. The fused rings in the matrix varied in size and nature from the fine separation shown in Fig.10 to the complete pool of eutectic shown in Fig.11, (the section was etched in a saturated solution of hydrogen sulphide in 0.1% sulphuric acid). The lake of ternary $Al_2Mg_3Zn_7$ shown in Fig.11 also contains a particle of $MnAl_6$ and one of Mg_2Si . The formation of fine rings similar to that shown in Fig.10 appears to be the first indication of overheating in the alloy and it is probable that this is the first stage in the formation of pools of fused material in the matrix.

In the 4:1:9, 4:1:10 and 4:1:11 alloys treated at 470°C. for one hour any ternary $Al_2Mg_3Zn_7$ that had not been taken into solution at this temperature was globularised without the formation of any thin films of eutectic which would render the material brittle. This also occurred during the solution treatment of alloy 5:1:12 at 450°C.

Discussions of results and conclusions

The excellent forging qualities of the materials throughout the whole range investigated suggests that alloys of similar compositions would be suitable for other hot working processes, as for instance the production of bars and sections by extrusion and of sheet and strip by rolling. Furthermore it would be expected that these methods of fabrication under industrial conditions would result in improved tensile strength and ductility.

Alloys for which the sum of magnesium and zinc contents did not exceed 12% were successfully heat treated by solution treatment for one hour at 495°C., followed by quenching in cold water, and ageing for 18 hours at 120°C. It was considered that for alloys of a higher alloy content, a temperature of 470°C. was preferable for solution treatment in order to reduce the risk of incipient fusion. For the 5:1:12 alloy the maximum temperature which could be used for solution treatment was 450°C. Values for 0.1% proof stress, ultimate stress and elongation for the 2:1:11 alloy in the fully heat treated condition were 37.0 tons/in.², 40.1 tons/in.² and 6% respectively, and for the 4:1:11 alloy, 40.4 tons/in.², 44.5 tons/in.² and 2% respectively. The proof and ultimate tensile strength values are of a high order but the elongation value of 2% is lower than the minimum of 5% specified for wrought aluminium alloys for use in aircraft structures, (see Specification D.T.D.363A). It is probable that alloys of similar composition produced under commercial conditions may show improved ductility on the alloys prepared at R.A.E. Further, by suitable modification of the heat treatment, an increase in ductility might be achieved with a relatively small loss in tensile strength. It would however be necessary to investigate the effect of varying the heat treatment, and in particular the temperature of ageing, on the stress-corrosion properties.

For a given percentage of magnesium, the value of 0.1% proof stress and ultimate tensile strength increase steadily to a maximum with increasing zinc content while the elongation value decreases. Similarly for a given percentage of zinc, increases in proof and tensile stresses and a decrease in elongation are obtained for an increase in the magnesium content.

It is suggested that for alloys in the ($\alpha + T$) field of the equilibrium diagram at the ageing temperature, (for which the hardening phase is probably $Al_2Mg_3Zn_7$), for a given set of conditions of solution treatment and ageing, the tensile strength obtained is a function of the product $[Mg][Zn]$, where $[Mg]$ and $[Zn]$ are the atomic percentages of free magnesium and of zinc respectively. Similarly, the tensile strength of alloys in the ($\alpha + MgZn_2$) field is considered to be a function of the product $[Mg][Zn]^2$.

Microscopical examination of the alloys in various conditions showed that the constitution of the alloys was very similar to that of higher purity materials. The presence of the $MgZn_2$ phase in alloys of relatively high alloy content did not lead to any difficulties provided that the solidus temperature of the alloy was not exceeded during solution heat treatment. Solution heat treatment at too high a temperature resulted in the formation of films and lakes of fused ternary constituent in the grain boundaries and intercrystalline weakness in the material.

In view of the promising tensile properties of the 2:1:11 alloy, further investigation of the properties of an alloy with 2 to 3% of magnesium and 10 to 11% of zinc is in hand. This work will include tensile, fatigue and stress corrosion tests on both rolled and extruded materials. Earlier work has indicated that the stress-corrosion properties of alloys of this type are improved by additions of chromium up to 0.5% which have little effect on the tensile properties provided that care is taken during casting to avoid marked segregation of chromium and manganese. It is suggested that stress-corrosion tests made on a series of wrought alloys of high tensile strength, both with and without chromium in each case, would yield interesting information.

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11. R.A.E. Note M.7773. "Microscopical examination of alloys of the Mg-Mn-Zn-Al type received from Dr. Hume-Rothery, Oxford." B.W. Mott. March 1943.
12. W. Bungardt and G. Schmitberger, Z.f. Metallkunde, Vol.35, No.2 February 1943, pp.47-55. "The stress corrosion properties of some Al-Mg-Zn alloys after ageing at elevated temperatures".

Attached:

Appendix I
Tables I - IIIa
Figs. -

(Drgs. Mat.2017-2021)
(Neg.No. M.7790)

Circulation:

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APPENDIXPreparation and casting of alloys

The analyses of the metals and temper metals used are given in Table I.

A covered Salamander crucible and an electric resistance furnace were used for the preparation and melting of all alloys.

About two thirds of the virgin aluminium required to make up a melt was melted first and raised to a temperature of about 800°C. The surface of the melt was then covered with Coveral flux, (Messrs. Foundry Services Ltd.), the titanium and manganese temper metals added and the temperature again adjusted to 800°C. Half the remaining aluminium was then introduced and after the temperature had been adjusted to about 750°C. addition of zinc followed. The balance of the aluminium was next introduced at about 740°C. and finally addition of the magnesium was made at a temperature of 700-720°C. by plunging in an inverted steel cup containing it, the cup being held beneath the surface of the melt until all the magnesium had been taken into solution. Aluminium chloride, amounting to approximately 0.1% of the total weight of the melt, was then added as a degasser. On the completion of the evolution of the gas and after the adjustment of the temperature to approximately 700°C., the metal was poured slowly into a chill mould, with sand cup head. The mould was preheated to a temperature of about 250°C., and inclined at an angle of about 30° to the vertical. As the mould was filled it was moved into an upright position, and additions of further metal were made to minimise piping.

Forging

From each cast billet, approximately 2 inches diameter by 6 inches long, two shorter billets were machined, approximately 1½ inches diameter by 2½ inches long, and also a thin slice for examination of the macro-structure.

After the internal soundness of each billet had been proved by radiographic examination, the billets were heated in an electric furnace fitted with an air circulating fan for about 18 hours at a temperature of 450°C., and were then upset under a 250-ton hydraulic press, the stroke of the press being in the direction of the original axis of the cast billet. After further heating at 450°C., the pressing operation was continued, and the resultant bar was drawn out from the billet in a direction at right angles to the original longitudinal axis. After further heating at 450°C., the bars were forged to ½ inch square section under a 10 cwt. pneumatic hammer.

TABLE IChemical composition of key alloys used

	Al.	Cu.	Fe.	Mg.	Mn.	Si.	Ti.	Zn.
Aluminium	Bal		0.28			0.18		
Magnesium	Nil	Trace	0.04	Bal	Trace	Trace		
Manganese-Aluminium	Bal		0.44		15.1	0.30		
Titanium-Aluminium	Bal		2.63			0.46	3.48	
Zinc			0.003					Bal

TABLE IINominal compositions of alloys investigated,
showing Mg-Mn-Zn contents

1:1:3	2:1:3	3:1:3	4:1:6	5:1:12
1:1:4	2:1:4	3:1:4	4:1:8	
1:1:5	2:1:5	3:1:5	4:1:9	
1:1:6	2:1:6	3:1:6	4:1:10	
1:1:8	2:1:8	3:1:8	4:1:11	
1:1:9	2:1:9	3:1:9		
1:1:10	2:1:10	3:1:10		
1:1:11	2:1:11	3:1:11		
		3:1:12		

TABLE IIIA

Chemical composition and tensile properties of alloys with 1% of magnesium

Nominal Composition		1:1:3	1:1:4	1:1:5	1:1:6	1:1:8	1:1:9	1:1:10	1:1:11
Composition obtained by chemical analysis	Mn	1.06	1.06	0.96	1.06	1.12	1.09	1.02	1.02
	Si	0.15	0.15	0.15	0.15	0.16	0.14	0.14	0.11
	Fe	0.31	0.32	0.31	0.32	0.31	0.34	0.31	0.42
	Zn	3.29	4.01	5.05	6.11	8.08	9.05	10.48	11.25
	Mg	1.10	1.09	1.07	1.18	1.02	0.86	1.04	1.06
Atomic %	Ti	0.04	0.04	0.04	0.04	0.05	0.05	0.05	0.05
	Al by dif.	94.04	93.33	92.42	91.14	89.26	88.47	86.96	86.09
Atomic % free Mg		1.38	1.70	2.16	2.63	3.52	3.97	4.65	5.01
Solution Heat Treatment Temp. °C.		495	495	495	495	495	495	470	470
Ageing Temp. °C.		120	120	120	120	120	120	120	120
Tens. Gauge Length. ins. F.L. 0.1% P.S. 0.2% P.S. U.T.S. E x 10 ⁻⁶ Elong. on 1/4" Red. of Area B.H.N. 10/1000/15 D.P.H.N. Hd/30	ins.	0.564	0.553	0.564	0.564	0.493	0.493	0.493	0.492
	ins.	2.0	2.0	2.0	2.0	2.0	2.0	2.0	2.0
	T/sq.in.	6.56	7.65	10.0	17.4	17.8	17.8	13.6	17.6
	T/sq.in.	8.28	11.4	16.4	23.4	25.2	24.1	29.6	29.2
	T/sq.in.	8.6	11.8	16.9	23.9	26.0	24.8	30.5	30.2
	T/sq.in.	16.8	19.8	22.7	28.3	29.8	28.6	33.2	32.6
	lb/sq.in.	9.8	9.9	9.8	9.95	9.8	9.9	9.8	9.7
	%	24	21	21	19	16	14	12	10
	%	50	35	43	37	30	29	19	16
	%	82	84	108	138	145	145	164	170
103		115	124	149	162	156	182	185	185

TABLE IIID

Chemical composition and tensile properties of alloys with 2% of magnesium

Nominal composition		2:1:3	2:1:4	2:1:5	2:1:6	2:1:8	2:1:9	2:1:10	2:1:11
Composition obtained by analysis	Mn	1.09	1.03	1.02	0.99	1.12	1.03	1.02	0.99
	Si	0.16	0.12	0.13	0.13	0.14	0.15	0.14	0.10
	Fe	0.32	0.34	0.34	0.31	0.34	0.36	0.34	0.39
	Zn	3.13	4.12	4.97	6.10	8.10	8.93	10.35	11.32
	Mg	1.96	1.85	2.26	2.03	2.19	2.01	2.05	1.83
	Ti	0.04	0.05	0.05	0.04	0.05	0.05	0.04	0.04
Al by diff.		93.30	92.49	91.23	90.40	88.06	87.41	86.06	85.53
Atomic % Atomic % free Mg	Zn	1.32	1.75	2.12	2.62	3.52	3.92	4.55	5.03
	Mg	1.91	1.88	2.43	2.11	2.28	2.07	2.14	1.98
Solution Heat Treatment Temp. °C. Ageing Temp. °C.		495 120	495 120	495 120	495 120	495 120	495 120	470 120	470 120
Dia. Gauge lgth. P.L. 0.1% P.S. 0.2% P.S. U.T.S. E x 10 ⁻⁶ Elong. on 1/4" Red. of Area B.H.N. 10/1000/15 D.P.H.N. H _D /30	ins.	0.563	0.564	0.565	0.564	0.494	0.494	0.493	0.493
	ins.	2.0	2.0	2.0	2.0	2.0	2.0	2.0	2.0
	T/sq.in.	6.66	12.7	18.3	21.0	22.9	22.9	22.3	19.5
	T/sq.in.	10.6	18.0	25.5	28.4	34.4	34.4	37.2	37.0
	T/sq.in.	11.0	18.7	26.3	29.2	35.4	35.4	38.1	38.6
	T/sq.in.	20.0	25.4	31.2	33.1	37.7	37.6	40.0	40.1
	lbs/sq.in.	9.75	9.65	9.8	9.8	10.0	9.8	9.6	9.7
	%	24	18	18	18	6	7	5	6
	%	48	39	35	28	8	11	10	13
		87	117	147	156	179	179	194	197
		118	127	159	176	204	206	210	217

Chemical composition and tensile properties of alloys with 3/4 of magnesium

Nominal Composition		3:1:3	3:1:4	3:1:5	3:1:6	3:1:8	3:1:9	3:1:10	3:1:11	3:1:12
Composition obtained by analysis	Mn	1.00	1.00	0.98	1.00	1.01	1.02	1.04	1.02	0.97
	Si	0.13	0.14	0.16	0.14	0.15	0.15	0.15	0.09	0.14
	Fe	0.31	0.34	0.36	0.31	0.34	0.31	0.31	0.36	0.31
	Zn	2.88	4.24	4.99	6.02	8.10	9.23	10.74	10.96	12.7
	Mg	2.86	3.00	3.13	3.03	2.81	3.11	3.04	2.85	3.12
Al by dif.	Ti	0.04	0.05	0.05	0.05	0.05	0.04	0.04	0.05	0.05
	Al by dif.	92.78	91.23	90.32	89.45	87.54	86.14	84.68	84.67	82.71
Atomic % Zn		1.22	1.79	2.13	2.59	3.52	4.09	4.75	4.85	5.70
Atomic % free Mg		3.01	3.16	3.27	3.23	3.00	3.40	3.32	3.21	3.58
Solution Heat Treatment Temp. °C.		495	495	495	495	495	495	470	470	495
Ageing Temp. °C.		120	120	120	120	120	120	120	120	120
Dia. Gauge lgth.	ins.	0.562	0.564	0.564	0.563	0.565	0.494	0.493	0.495	0.493
	ins.	2.0	2.0	2.0	2.0	2.0	2.0	2.0	2.0	2.0
P.L.	T/sq.in.	7.65	18.1	18.8	26.3 ^a	19.7	25.4	26.2	23.7	22.8
0.1% P.S.	T/sq.in.	11.3	23.5	27.8	31.7 ^a	36.8 ^b	39.3	40.4	40.1	39.8
0.2% P.S.	T/sq.in.	11.9	24.5	28.9 ^b	32.8 ^b	N.D.	40.6	41.7	41.5	41.5
U.T.S.	T/sq.in.	22.8	30.4	33.2	36.4	39.8	43.0	43.8	43.8	43.8
E x 10 ⁻⁶	lb/sq.in.	9.75	9.9	9.9	9.85 ^a	10.05	9.6	9.6	9.7	9.7
Elong. on 1/4"	%	22	15	10	12.5	2	2	3	2	2
Red. of Area	%	43	32	20.	20	3	N.D. ^c	3	2	2
E.H.N.10/1000/15		96	140	151	168	184	203	211	209	211
D.P.H.N. HD/30		120	160	187	195	210	226	232	229	231

a. Evidence of extensometer slip. L.P. value very unreliable. P.S. and E fairly accurate.
 b. Obtained by extrapolation.
 c. Fracture occurred in end threads.

d. Failed in threads. Load extension curve irregular above 15 T/sq.in. E obtained from straight part of load extension curve.
 U.T.S. and failing load divided by C.S. area of gauge length.

TABLE IIIA

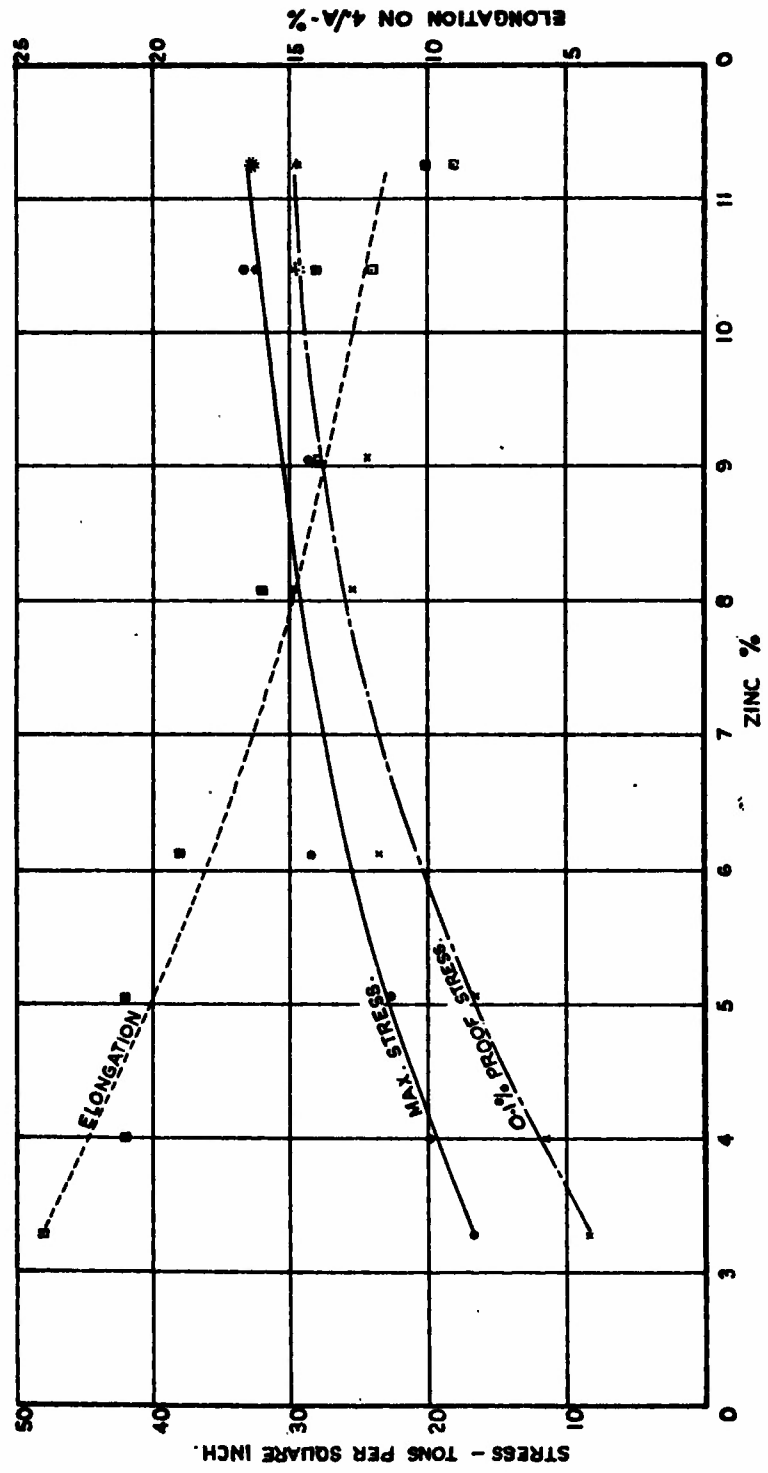
Chemical composition and tensile properties of alloys with 4 and 5% of magnesium

Nominal Composition		4:1:6	4:1:8	4:1:9	4:1:10	4:1:11	5:1:12
Composition obtained by analysis	Mn	0.98	1.09	1.06	1.01	0.98	1.05
	Si	0.14	0.15	0.15	0.10	0.10	0.14
	Fe	0.34	0.34	0.31	0.34	0.34	0.34
	Zn	5.95	8.39	9.02	9.92	11.35	11.1
	Mg	3.95	4.07	4.26	3.82	4.08	4.55
Al by Dif.		0.06	0.05	0.04	0.05	0.05	0.06
		88.58	85.91	85.16	84.76	83.10	82.76
Atomic %							
Zn		2.54	3.67	3.95	4.35	5.03	4.20
Atomic % free Mg		4.27	4.47	4.71	4.53	4.67	5.13
Solution Heat Treatment Temp. °C.		495	495	470	470	470	495
Aging Temp. °C.		120	120	120	120	120	120
Dia. Gauge Lgth. P.L. 0.1% P.S. 0.2% P.S. U.T.S. E x 10 ⁻⁶	ins.	0.564	0.169	0.493	0.493	0.493	0.494
	ins.	2.0	1.0	2.0	2.0	2.0	2.0
	T/sq.in.	15.4	21.9	21.1	21.5	21.5	21.5
	T/sq.in.	33.2	37.8	38.5	38.0	39.9	40.0
	T/sq.in.	34.4 ^a	39.5	39.9	41.2	41.5	41.7
	T/sq.in.	37.9	42.6	42.6	44.0	44.5	43.7
	T/sq.in.	10.1	9.7	9.8	9.8	9.8	9.8
	lbs/sq.in.	6	< 1	2	2	2	2
	Elong. on 1/4" %	8	N.D. ^b	4	4	4	4
	Red. of Area %	177	198	202	210	220	197
B.H.N. 10/1000/15		201	215	226	232	230	180
D.P.H.N. Hd/50		201	217	226	228	228	-

a. Obtained by extrapolation

b. Fracture occurred in end threads

c. Failed on shoulder. Load extension curve irregular above 64 T/sq.in. E obtained from straight part of load extension curve. U.T.S. and failing load divided by C.S. area of gauge length.



SOLUTION TREATMENT
TEMP. 1 HOUR AT

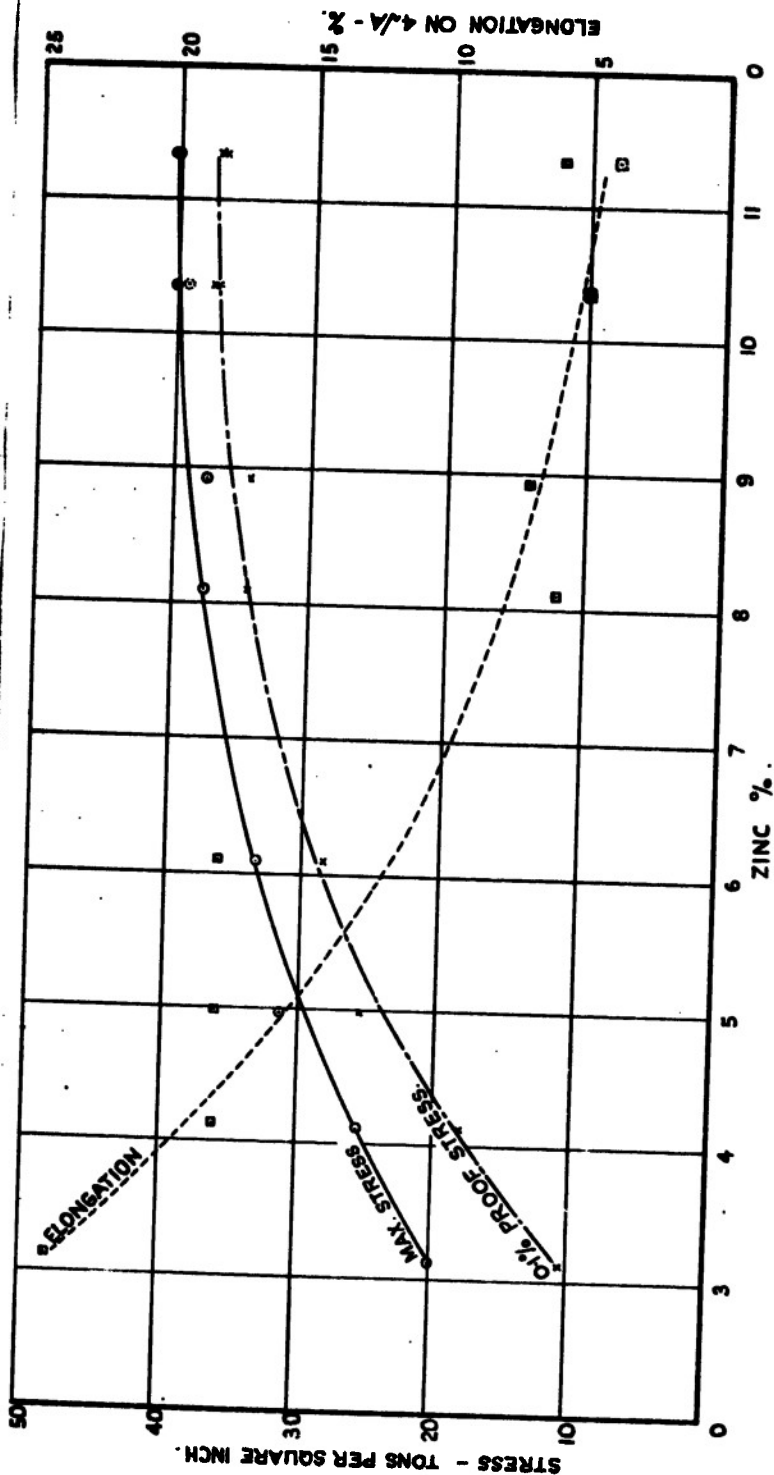
495°C.

470°C.

1% MAGNESIUM.

RELATION BETWEEN TENSILE PROPERTIES AND ZINC CONTENT
FOR ALLOYS CONTAINING 1% OF MAGNESIUM. (17)

DUPLICATE TEST RESULTS SHOWN THUS □ □ □ □



SOLUTION TREATMENT
TEMP. 1 HOUR AT

495°C

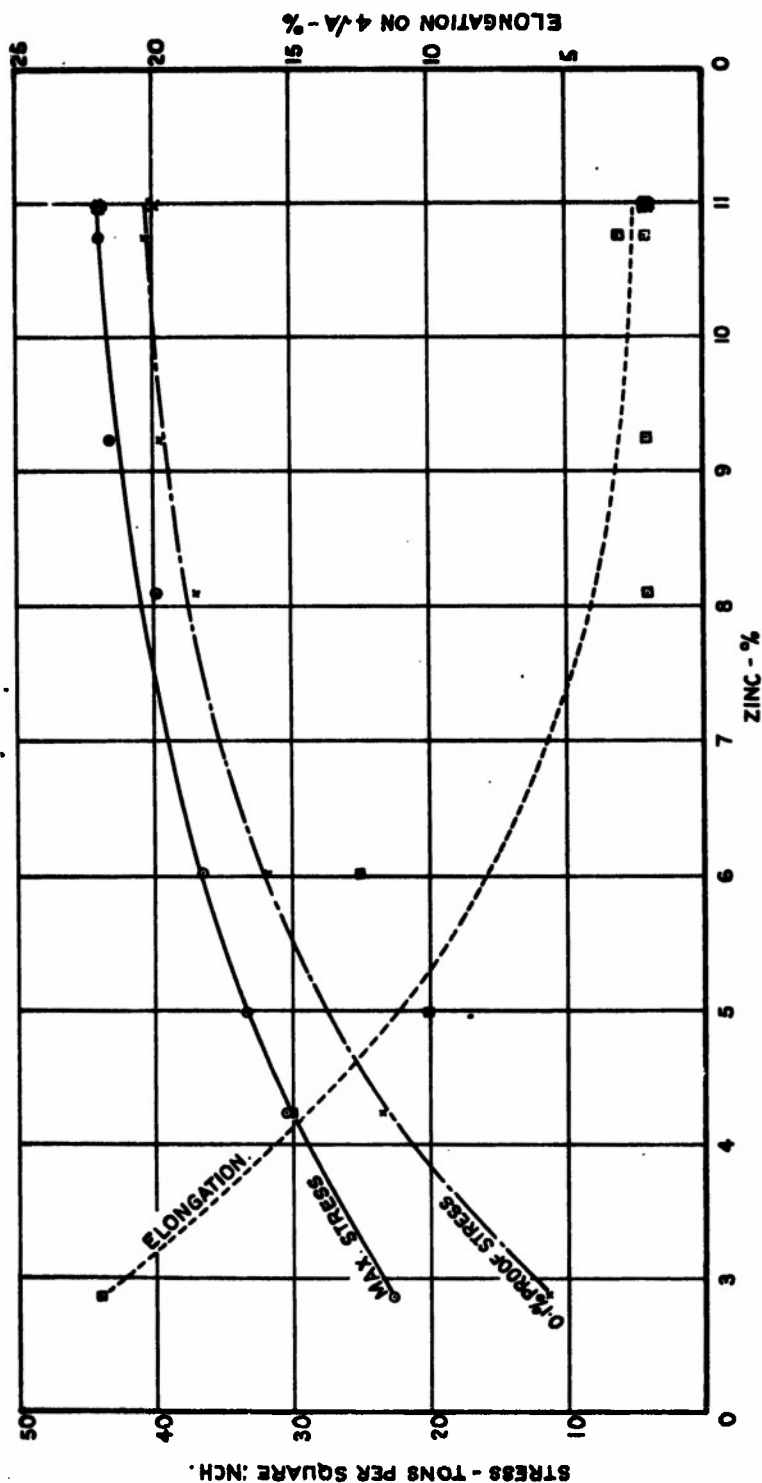
DUPLICATE TEST RESULTS SHOWN THUS ○○○

470°C

2% MAGNESIUM.

RELATION BETWEEN TENSILE PROPERTIES AND ZINC CONTENT
FOR ALLOYS CONTAINING 2% OF MAGNESIUM.

18



SOLUTION TREATMENT

TEMP. 1 HOUR AT

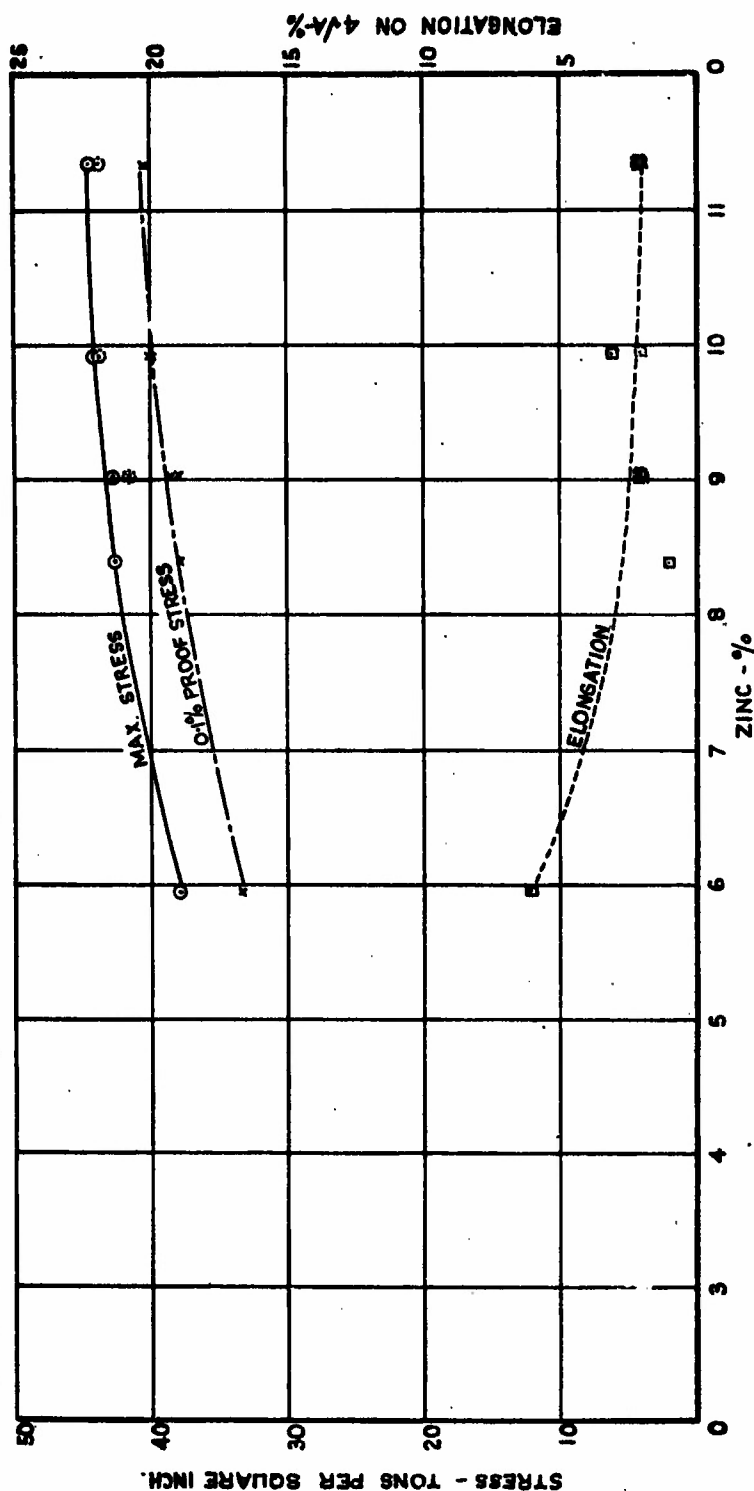
495°C

DUPLICATE TEST RESULTS SHOWN THUS

470°C 3% MAGNESIUM.

RELATION BETWEEN TENSILE PROPERTIES AND ZINC CONTENT
FOR ALLOYS CONTAINING 3% OF MAGNESIUM.

19



SOLUTION TREATMENT
TEMP. 1 HOUR AT
DUPLICATE TEST RESULTS SHOWN THUS ⊗ ⊗ X

495°C. 470°C. 4% MAGNESIUM.

RELATION BETWEEN TENSILE PROPERTIES AND ZINC CONTENT
FOR ALLOYS CONTAINING 4% OF MAGNESIUM.

(20)

FIG. 1 E

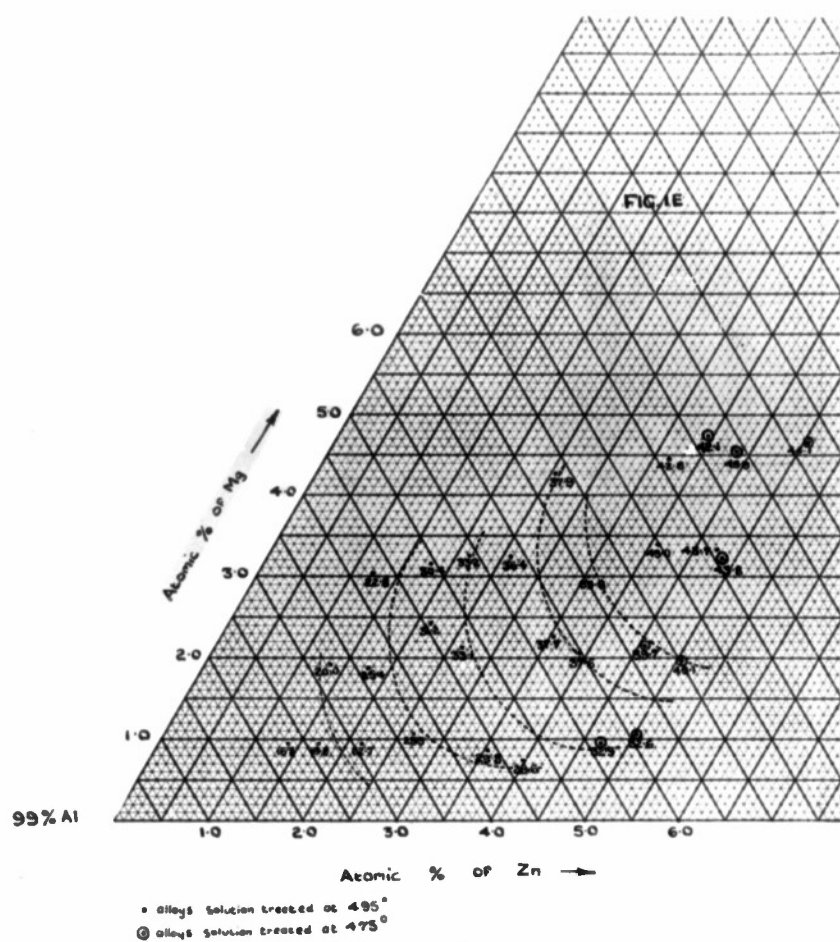
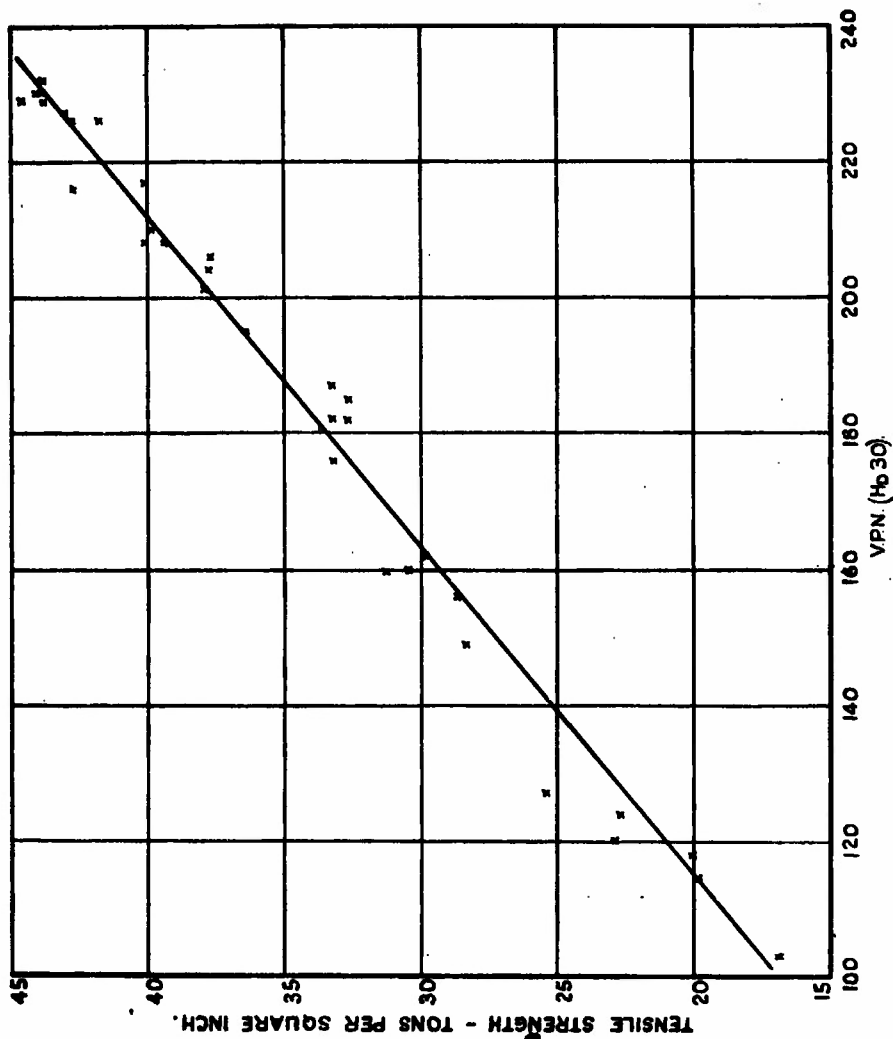


FIG. 1 E

(21)



RELATION BETWEEN TENSILE STRENGTH AND HARDNESS OF ALLOYS.

-22-

FIG. 3

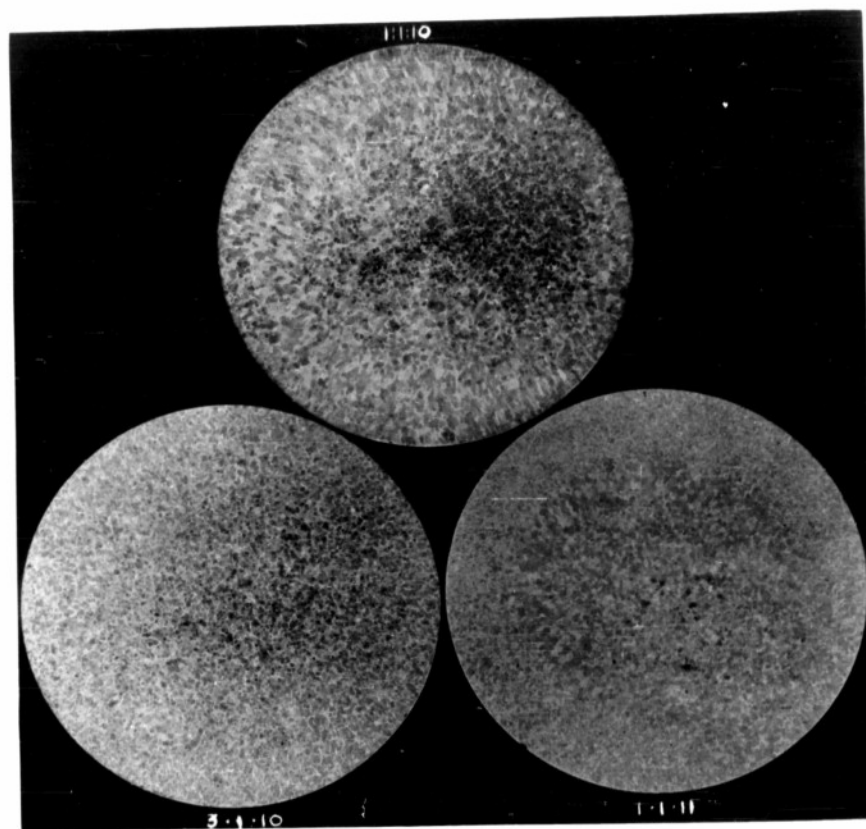


FIG. 3

X 1 1/2

RAE NEG. NO. 71643 /46

(23)

FIG 4, 5, 6 & 7

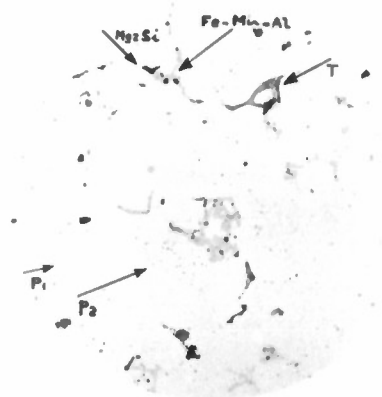


FIG. 4

X 400



FIG. 5

X 400

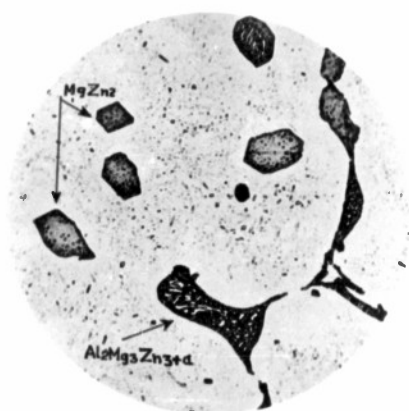


FIG. 6

X 750

FIG. 7

X 750

RAE NEG. NO. 71684/46

24

FIG 8, 9, 10 & 11

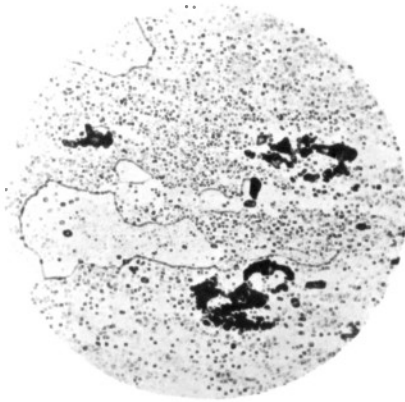


FIG. 8

X 750

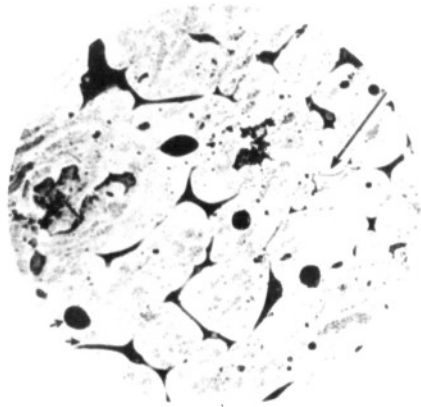


FIG. 9

X 200

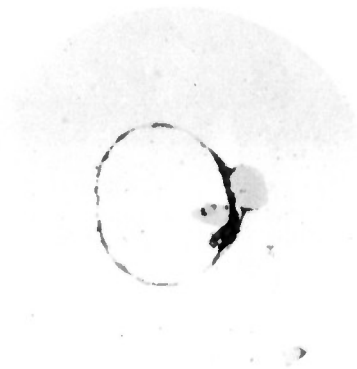


FIG. 10

X 3000



FIG. 11

X 2500

R.A.E. NEG. NO. 71645/46



REEL - C

3 4 7

A.T.I.

1 0 1 3 7

RESTRICTED

TITLE: The Relation Between the Tensile Properties and the Constitution of Aluminum rich Aluminum-Magnesium-Manganese-Zinc Alloys

AUTHOR(S) : Ross, A. H., Mott, B. W.

ORIG. AGENCY : Royal Aircraft Establishment, Farnborough, Hants

PUBLISHED BY : (Same)

ATP- 10137

(None)

ORD. AGENCY ED.

MIT-7

PUBLISHED AGENCY ED.

(Same)

DATE	EDIC. CLASS.	COUNTRY	LANGUAGE	PAGE	ILLUSTRATIONS
Aug '48	Restr.	Gt. Brit.	English	25	photos, tables, graphs

ABSTRACT:

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DIVISION: Materials (8)

SECTION: Aluminum and Alloys (10)

SUBJECT HEADINGS: Aluminum alloys - Physical properties (10588.3); Aluminum alloys - Strength (10589.6)

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TITLE: The Relation Between the Tensile Properties and the Compositions of Aluminum-rich Aluminum-Magnesium-Manganese-Zinc Alloys

AUTHOR(S): Ross, A. H., Mott, B. W.

ORIG. AGENCY: Royal Aircraft Establishment, Farnborough, Hants

PUBLISHED BY: (Same)

AT- 10137

(None)

ORIG. AGENCY ID.

MUT-7

PUBLISHED AGENCY ID.

(Same)

DATE	DOC. CLASS.	COUNTRY	LANGUAGE	PAGES	ILLUSTRATIONS
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Record Summary: AVIA 6/13106

Title: Tensile properties and constitution of aluminium-rich aluminium-magnesium-manganese-zinc alloys

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